

EFFECT OF CONCURRENT PHASE TRANSFORMATION ON PLASTIC DEFORMATION BEHAVIOUR OF Al-Cu ALLOYS

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for the Degree of
MASTER OF TECHNOLOGY**

**By
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CERTIFICATE

This is to certify that this work 'Effect of Concurrent Phase Transformation on Plastic Deformation Behaviour of Al-Cu Alloys' has been carried out by Mr. B.P. Rao under my supervision and it has not been submitted elsewhere for a degree.


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THESIS APPROVAL OFFICE
This thesis has been approved
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ABSTRACT

Commercial Aluminium melted ~~at 900°C~~, to which commercial copper in the range (1-4 percent Cu) was added to prepare the alloys. Homogenization was carried out at 520°C for 20 hours and hot rolled at 400°C. Tensile and hot hardness samples were made and annealed at 520°C for 4 hours.

Hot indentation tests were carried out at 360°C, 400°C, 450°C, and 500°C for each composition by varying testing periods upto 2 hours. Similar tests were carried in tensile samples for each composition tested during $\alpha + \theta \rightarrow \alpha$ (single phase) and in two phase regions by varying testing periods from 5 to 120 minutes in the strain rate range of 10^{-3} min^{-1} to 10^{-1} min^{-1} .

Hot indentation tests show that hardness drops continuously reaching a minima over a period of 1 hour and again rises slightly with time. The peak SRS values increase with percentage of copper and remain in superplastic range ($m \geq .3$) over a period of 2 hours in 3 percent and 4 percent Cu during $\alpha + \theta \rightarrow \alpha$. These values in comparison with Aluminium are far better. The peak SRS in single and two phase regions although not differ much in 1 percent and 2 percent Cu but in case of 3 and 4 percent ~~in~~ the difference is quite clear.

The drop in hardness over a period of 1 hr. reaching minima attribute that additional softening processes might occur apart from general softening, for which diffusional flow mechanisms like Nabarro or Coble creep have to be invoked. This helps to homogenize the material during softening. From tensile tests for higher holding times the increase in volume fraction continues dissolution reaction for long periods,

keeping $m \geq .3$. It is proposed that during dissolution reaction there is a transitory development of fine grain size (1 - 10 μ) which extends even to homogenization periods because of Nabarro or Coble creep mechanism operating which predict high 'm' values. Thus this model supports the previous observations reported in Sn-Sb system [17].

CHAPTER - I

INTRODUCTION

The plastic deformation behaviour of metals and alloys has been studied extensively with a view to understand their processing behaviour. Metal-processing technologists are interested in large ductility (prolonged neck free deformation) and lower deformation load to economize the processing. Their ideal is 'glassy' behaviour where material would show extensive ductility under very low stresses.

Characterization of Plastic Deformation

Plastic flow in metals is characterized by a flow equation:

$$\sigma = K \epsilon^n \dot{\epsilon}^m \quad (1)$$

where σ is the stress to cause the plastic deformation, K is a material constant, ϵ is the plastic strain, $\dot{\epsilon}$ plastic strain rate, 'n' is strain hardening index and 'm' is a strain rate hardening index or strain rate sensitivity (SRS).

Factors Affecting the Plastic flow

The factors which affect the plastic flow of metals are as follows:

1. Temperature
2. Microstructure
3. Phase Transformation

Temperature

It is well known that increasing the temperature of a metal helps to lower down the deformation loads. This is achieved through progressive loss of yield strength of the metal with increasing temperatures. Further, at temperature higher than $0.5 T_m$, where T_m is the melting temperature of the metal expressed in $^{\circ}\text{K}$, recovery processes are fast enough to overcome the phenomenon of strain hardening and help retain deformability of the metal even after progressively increasing deformation.

At low temperatures (~~#~~^{<<} $0.5 T_m$), the index 'm' in the equation (1) tends to be zero and the plastic flow behaviour is governed by:

$$\sigma = K' \epsilon^n \quad (2)$$

Strain hardening index, 'n' becomes a measure of ductility in the material.

At high temperatures ($\geq 0.5 T_m$), where recovery processes are predominant, the strain hardening index, n $\rightarrow 0$ and the flow behaviour is now governed by the equation

$$\sigma = K'' \dot{\epsilon}^m \quad (3)$$

SRS (m) becomes a measure of ductility (see Appendix I) in the material at these temperatures.

Microstructure

Generally, a single phase material is easier to deform plastically. Presence of two or more than two phases, especially when one of the phases is very strong is difficult to deform.

In recent times, it has been established (~~Ref~~ [26]) that one can induce large scale ductility ($\geq 300\%$) in metals by obtaining a very fine grained ($< 10 \mu$) microstructure and carrying out deformation at temperatures higher than $0.5 T_m$. It has been designated as 'Superplasticity'.

Extensive studies on fine grained metals and alloys have revealed that the main cause of 'superplastic deformation' in these materials is the high SRS values ($m \geq 0.3$) shown by them. There is also, lowering of deformation loads during the superplastic flow. It has been attributed to such internal mechanisms as, (i) Grain Boundary Sliding and (ii) Nabarro-Herring or Coble Creep flow, which need lower operating stresses as compared to the usual dislocation controlled plastic flow.

Phase Transformation

Plastic deformation behaviour is drastically

altered if there is a phase transformation reaction taking place while the deformation is being carried out. Following effects have been reported in this regard:

- (i) General weakening (lowering of yield strength) of material ~~(Ref.~~ [10,11].
- (ii) Higher SRS values i.e. ($m > .3$) [17].
- (iii) Increased ductility ~~(Ref.~~ [14,15].

The general weakening of materials during a transformation reaction would help reduce deformation loads during processing. The high SRS values should contribute to large ductility values. Further, it has been shown that superplastic flow could be achieved by repeated cycling through transformation temperature range.

Scope of Present Investigation

Most of the studies on the effects of concurrent transformation have been confined to steels. It is probably, because of their extensive commercial usage. There are no reported studies on Al-Cu system which is one of the systems of commercial importance.

In Al-Cu system (Fig.1) at small percentages of copper ($\leq 5\%$), there is a transformation reaction designated as:



where α is a solid solution and θ is an intermetallic compound, CuAl_2 . It is with a view to exploit this reaction towards processing ease, the present investigation has been carried out. The objectives of the present investigation are:

1. To observe whether $\alpha \rightleftharpoons \alpha + \theta$ reaction would lead to softening of Al-Cu alloys during the reaction.
2. To study the effect of the above reaction on the SRS values.
3. To understand the effect of holding time and increasing copper percentage on the SRS?

These are the limited objectives of the present investigations.

To achieve these objectives two kinds of tests were performed, viz. (i) Hot Hardness test and (ii) Differential strain rate test [29].

CHAPTER - II

LITERATURE REVIEW

In the introduction (i.e. Chapter I) it has been very briefly mentioned that the concurrent transformation affects (1) Strength (2) Ductility (3) Strain Rate Sensitivity and (4) Creep Behaviour in a material. The purpose of this Chapter is to review some of the reported studies and the explanations offered for the observed effects.

2.1 Effect on Strength

The earliest reported observations are by Sauveur [1] on Iron which was subjected to plastic deformation by twisting it in a temperature gradient. The material showed regions of easy twisting (i.e. regions of low strength) where temperature in the sample was close to transformation temperature. Similar observations were made by E. Scheil [2] and Koref [3]. They termed the effect as 'amorphous plasticity'. There was no proper explanation for the observed softening.

Bochvar studied the transformation effects on strength in terms of indentation hardness [4]. He showed that Al-Mg containing 5 % Mg was softer than alloys containing 3 percent or 7 percent Mg when tested at 300°C [5]. The softening was attributed to the proximity of the alloy containing 5 percent Mg to the solubility limit of Magnesium

in Aluminium at 300°C. In other words 'solution and precipitation of mutual phases' ($\alpha \rightarrow \alpha + \beta$ reaction) occurred in the 5 percent Mg alloy while hardness was being measured and the same was absent (or less pronounced) in the alloys of other compositions. The loss of strength was qualitatively attributed to high diffusion rates and fine dispersion of phases.

Lowering of hardness values during the transformation were also reported in steels [6] Magnesium alloys [7] and nickel alloys [8]. Most of the hot hardness observations indicate weakening of materials near transformation temperatures. However, there are few exceptional cases of strengthening reported by Kornilov and Co-workers [9].

Systematic investigations on Fe + 0.2 percent carbon steel were carried out by De Jong and Rathenau [10]. They have observed weakening of the steel during $\alpha \rightarrow \gamma$ allotropic transformation while twist deformation was carried out. They have attributed the weakening to internal stresses developed during the transformation process due to differences in specific volumes of the parent and the product phases.

Clinard and Sherby [11] attribute the weakening of Armo Iron, to the generation of excess interstitials or vacancies during the process. In other words they have suggested rapid diffusion during the transformation.

L Lozinsky and Simeonova [12] attribute weakening in Armco Iron to the loss of coherency of the crystal lattice.

Underwood [13] has proposed that there is a general weakening of atomic bonds during any kind of phase transformation and therefore weakening is always expected if the material is deformed during the concurrent transformation.

2.2 Effect on Ductility and Strain-Rate Sensitivity

Many workers [14,15,16] have shown that neck free elongations of greater than 300 percent could be achieved by temperature cycling around the transformation temperatures. They have attributed these large ductilities to high Strain Rate Sensitivity ($m \approx 1$) obtained during transformation. Most of the conclusions regarding high strain rate sensitivity were indirect since the tests were not Isothermal in nature. In order to verify whether m values are high ($m \approx 1$) during the transformation, Isothermal tests were conducted by Vaidya and Agarwal [17] in Sn-Sb system. They carried out tests at temperatures where dissolution reaction involving $\alpha + \beta \rightarrow \alpha$ took place. The results are given in Fig. 2. It clearly shows that ' m ' values are between 0.5 to .8 during the period when the transformation is occurring. When the transformation goes to completion the ' m ' values fall below 0.25. The activation energy for deformation is nearly

half that of self diffusion of tin.

They have proposed that during transformation, owing to volume difference between α and β phases internal stresses are developed. These in turn lead to sub grain formation (subgrain size 10 μm). The fine sub grain size helps faster diffusion and in turn leads to high SRS values due to Nabarro or Coble type of viscous flow mechanisms.

Yoder and Weiss [18] recently carried out Isothermal tests in Eutectoid steel while pearlite changes to Austenite. They however failed to detect high 'm' values during the Isothermal testing. This is in contrast with Vaidya and Agarwal [17].study.

2.3 Effect on Creep Behaviour

The first observation on creep behaviour was made by Pfiel on 80 Ni-20Cr alloy [19] with varying amounts of Titanium. He found that creep resistance near solid solubility limit (775°C) increased rapidly, but with further addition of Ti it decreased, because of precipitation effects.

A study by Guy and Pavelick [20] was made in tin alloys containing Antimony in solid solution. They found an anomaly in creep rate near the solid solubility limit. In view of the recognized importance of grain boundary shearing at high temperatures during creep, they attributed the observed anomaly in creep rate to be associated with

grain boundaries.

Investigations by Porter and Rosenthal were made on Eutectoid steel [21] transforming to pearlite, bainite and martensite, under dead-weight tensile loading during Isothermal transformation. It appears that a threshold stress is associated with plastic deformation during transformation. In case of pearlite and bainite transformations increased rates of transformation occur at same threshold stress, while in case of Martensite transformation the gross plastic deformation occurs at very low stress as shown in Figs. (3, 4 and 5).

They attribute the increase in plastic deformation due to dislocation pile ups at grain boundaries and other barriers produce stress fields, resulting in increase of nucleation rate. When the coherency between parent austenite and nucleus is lost the advancing interface will act as sink for these piled up dislocations, so a threshold stress is necessary to move **them** out in order to prolong plastic deformation.

Similar studies were made on creep behaviour in Al-Mg alloys [22], Iron during $\alpha \rightarrow \gamma$ transformation [23, 24] and Indium alloys [25] respectively.

From the literature survey, it is observed that there exists no study of transformation effects on the

deformation behaviour of such an important system as Al-Cu in the range of 1 - 4 percent Cu . There are some studies related to micrograin plasticity [26, 27 and 28]. These are not concerned with the transformation effects.

CHAPTER - III

EXPERIMENTAL PROCEDURE

3.1 Sample Preparation

Commercial Aluminium (99.9 percent) was melted in clay graphite crucibles. ~~at 900°C~~. Commercial copper (99.9 percent) was added in composition by weight from 1 to 4. percent and casted in C.I. 'L' shaped moulds. The alloys were homogenized at 520°C for 20 hours and hot rolled at 400°C into thin strips. Tensile samples were made as per dimensions shown in Fig. (6). The samples were annealed at 520°C for 4 hours before testing.

3.2 Mechanical Testing

Tensile tests were carried out on Instron machine (Floor model/TT-C standard), by incremental strain rate method ranging from .005 cm/min - 5 cm/min. The test temperature were selected below and above the solvus line and a high temperature furnace capable of reaching 3000°F (supplied by Instron Co.) provided with a temperature controller (controlled within $\pm 4^{\circ}\text{C}$) was used. The samples were tested by introducing them when test temperature was reached and waited till the sample attained the required temperature measured by Alumel-Chromel thermocouple.

3.3 Method of Testing

The procedure suggested by Backofen et al [29] i.e. incremental step change in cross head speed after a steady state is reached for a particular cross head speed was adopted. Strain rate sensitivity Index 'm' was calculated according to the formula

$$m = \log(P_1/P) / \log(V_1/V) \quad \text{where}$$

P_1 = Maximum load at a cross head speed V_1

P = Maximum load at a cross head speed V

Stress and strain rate were calculated from instantaneous specimen length assuming uniform elongation. Since the measurements were done by increasing cross head speed, the method will introduce minor error until the high speeds were reached. The data for specimens showing necking on load vs time chart were neglected because most of the strain will occur at high strain rates. The furnace was controlled in such a way so that time interval between ON and OFF cycle was same and the time of testing the specimens ranged from 5 to 120 min. after attaining temperatures required.

3.4 Hot Hardness Tests

(a) Material Preparation

The samples were made in the same way as tensile specimens, except circular samples of 3 cm x 2.54 cm were prepared.

(b) Testing Method

The experimental set up is shown as in Fig. (7). It mainly consists of a furnace capable of reaching 600°C , a controller to control temperature within $\pm 4^{\circ}\text{C}$, and a dimmerstat (0-280 V) and a Alumel-Chromel thermocouple.

The furnace is placed on the base of Brinell-Hardness testing machine, provided with a central screw which can move up and down by turning base wheel without disturbing the furnace. The sample was placed on the platform of central screw provided for the purpose. A thermocouple was placed on the specimen surface as shown in the Fig. (7). The furnace, controller and Dimmerstat are joined in series.

Once the test temperature was attained in the furnace the sample was inserted and waited till the specimen also reached the testing temperature. Initial hardness was taken by applying a load 300 kg for 30 sec. At time intervals adjusted shown in the tables the impressions were taken and measured by Brinell Microscope at the end of testing. Hot hardness **B.H.N.** was calculated from the formula

$$\text{B.H.N.} = \frac{2P}{\pi D(D - \sqrt{D^2 - d^2})} \quad \text{where}$$

P = load applied in kg

D = diameter of the ball in mm (i.e. 10 mm).

d = diameter of the impression in mm

CHAPTER - IV

RESULTS

4.1 Variation of Indentation Hardness with Time at Constant Temperature

Tables (1-4) and Figs.(8-11) show that for each alloy, hardness drops continuously with time till it reaches a minima, and then rises again. The subsequent rise in the hardness depends upon the temperature of testing and the composition of the alloy. The minima in the hardness values lie in the time range of 30-75 min.

4.2 Comparison of Peak Strain Rate Sensitivity vs time in Commercial Aluminium and Al-Cu Alloys

The Figs (12-14) and table 5 show that peak 'm' values of 2-4 percent Cu in comparison with peak 'm' values of commercial aluminium tested at 450°C, 500°C and 520°C are higher and are well in superplastic range. i.e. ($m > .3$). It is also clear from these figures that peak 'm' values in case of 2-4 percent Cu alloys can be retained over a longer time of testing upto two hours.

4.3 Variation of Peak SRS with Time at a Constant Temperature in Single Phase Region

The SRS values vary with the applied strain rates. In all the cases a peak SRS value exists at a given

holding time of testing.

Fig. (15) and Table (6) indicate that the peak SRS value, for a given composition seems to drop with the increasing holding times. The peak strain-rate sensitivity values increases with percentage of copper (1-4 percent Cu) and values as high as 0.8 have been observed. It is clear from this plot (Fig.15) that as the percentage of Copper increases (from 1-4 percent) the values of peak 'm' remain above 0.3 i.e. $m > .3$ even for tests conducted upto two hours.

4.4 Comparison of Peak Strain Rate Sensitivity in Single and Two Phase Region

From the Table (7) it is clear that peak 'm' values when tested in single and two phase regions for lower percentages of copper i.e. 1 and 2 percent, there seems to be not much of difference. In case of 3 percent and 4 percent Cu there is a definite difference in peak 'm' values, clearly showing that better peak 'm' values can be obtained in case of single phase regions.

TABLE 1EFFECT OF TIME ON HOT HARDNESS OF Al-1% Cu ALLOY

Time in (Min.)	360°C B.H.N.	400°C B.H.N.	450°C B.H.N.	500°C B.H.N.
0	32	15.82	9.55	7.57
10	14.25	-	-	-
20	12.5	11.57	-	-
30	13.08	-	8	5.94
40	11.57	9.55	-	-
50	11.1	-	-	-
60	11.57	9.55	5.63	4.74
90	13.08	9.55	7.26	5.94
120	13.08	12	7.26	5.94

TABLE 2EFFECT OF TIME ON HOT HARDNESS OF Al - 2% Cu ALLOY

Time in (Min.)	360°C B.H.N.	400°C B.H.N.	450°C B.H.N.	500°C B.H.N.
0	28.08	16.18	11.57	7.57
10	21.7	-	-	-
20	18.54	10.67	-	-
30	16.75	-	9.55	4.74
40	16.18	10.26	-	-
50	14.8	-	-	-
60	14.25	9.18	7.92	3.85
90	16.75	13.08	9.55	4.94
120	16.75	13.08	9.55	5.94

TABLE 3

EFFECT OF TIME ON HOT HARDNESS OF Al - 3% Cu ALLOY

Time in (Min.)	360°C B.H.N.	400°C B.H.N.	450°C B.H.N.	500°C B.H.N.
0	19.48	15.52	9.55	7.57
10	14.8	-	-	-
20	11.1	11.57	-	-
30	11.1	-	6.67	4.74
40	12.01	11.57	-	-
50	12.01	-	-	-
60	12.01	9.89	6.67	4.74
90	13	12.6	7.92	5.94
120	13	12.6	7.92	5.94

TABLE 4EFFECT OF TIME ON HOT HARDNESS OF Al - 4% Cu ALLOY

Time in (Min.)	360°C B.H.N.	400°C B.H.N.	450°C B.H.N.	500°C B.H.N.
0	20.53	14.8	11.57	9.93
10	13.64	-	-	-
20	12.48	11.7	9.55	-
30	12.01	-	-	5.94
40	12.01	11.7	7.92	-
50	-	-	-	-
60	11.57	9.55	7.92	4.74
90	12.4	12	7.92	4.74
120	12.4	12	9.55	7.57

COMPARISON OF PEAK STRAIN RATE SENSITIVITY VS TIME IN COMMERCIAL ALUMINIUM AND
Al-Cu ALLOYS

450°C				500°C				520°C			
Time of holding (min)	Peak 'm' (2%Cu)	Peak 'm' Al	Time of holding (min)	Peak 'm' (3%Cu)	Peak 'm' Al	Time of holding (min)	Peak 'm' (4%Cu)	Peak 'm' Al	Time of holding (min)	Peak 'm' (4%Cu)	Peak 'm' Al
-	-	-	-	-	-	-	-	-	5	.746	.3479
-	-	-	-	-	-	-	-	-	8	.672	.2141
10	.45	.2	10	.321	.621	-	-	-	-	-	-
-	-	-	-	-	-	15	.799	.1926	-	-	-
18	.3626	.142	-	-	-	-	-	-	-	-	-
-	-	-	20	.73	.1332	-	-	-	-	-	-
30	.304	.273	-	-	-	30	.613	.1635	-	-	-
50	.170	.206	50	.341	.168	-	-	-	-	-	-
75	.247	.152	75	.43	.243	75	.3326	.1248	-	-	-
120	.170	.1255	120	.343	.241	120	.4671	.1844	-	-	-

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VARIATION OF PEAK SRS WITH TIME AT A CONSTANT TEMPERATURE IN SINGLE PHASE REGION

430°C (1 % Cu)		(450°C (2 % Cu)		500°C (3 % Cu)		520°C (4 % Cu)	
Time of holding (min.)	Peak 'm'	Time of holding (min.)	Peak 'm'	Time of holding (min.)	Peak 'm'	Time of holding (min.)	Peak 'm'
-	-	-	-	-	-	5	.746
8	.214	-	-	-	-	8	.672
-	-	10	.45	10	.321	-	-
-	-	-	-	-	-	15	.79
-	-	18	.3626	-	-	-	-
-	-	-	-	20	.73	-	-
21	.131	-	-	-	-	-	-
-	-	30	.304	-	-	30	.613
35	.148	-	-	-	-	-	-
50	.148	50	.170	50	.341	-	-
75	.2479	75	.247	75	.43	75	.3326
120	.148	120	.170	120	.343	120	.4671

TABLE 7

COMPARISON OF PEAK STRAIN RATE SENSITIVITY IN SINGLE
AND TWO PHASE REGION

Composition % Cu	Peak 'm' value in two phase region	Peak 'm' in single phase region
1	0.39	0.21
2	0.55	0.45
3	0.55	0.73
4	0.38	0.80

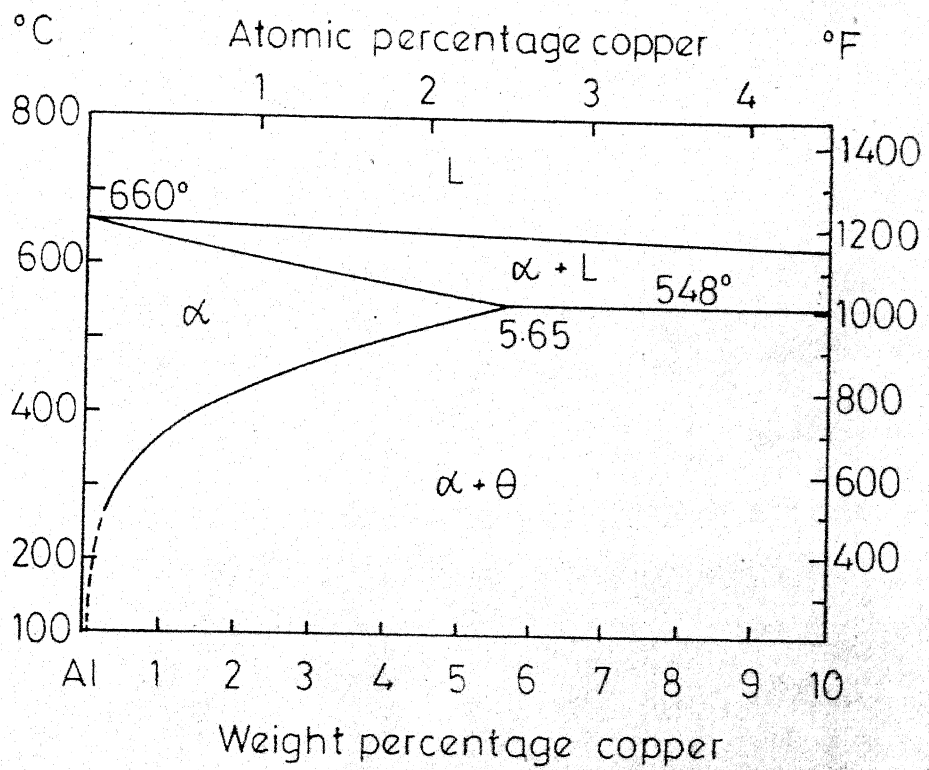


Fig. 1 - Phase diagram of Al-Cu system at aluminium end.

Test temperature 210°C

- 10 min
- 15 min
- △ 40 min
- 45 min

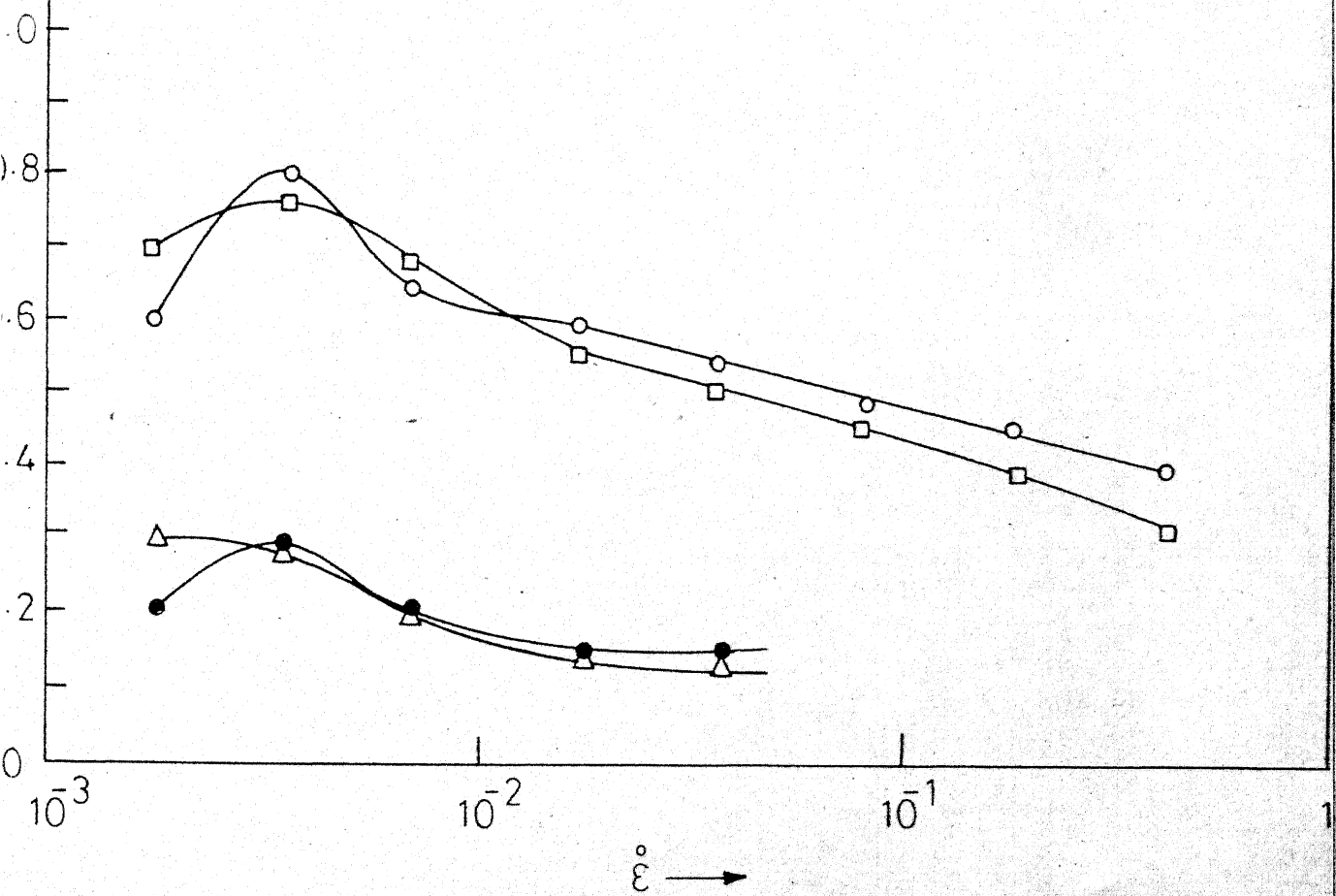


Fig. 2 - m vs. $\log \dot{\epsilon}$, Sb 6 %

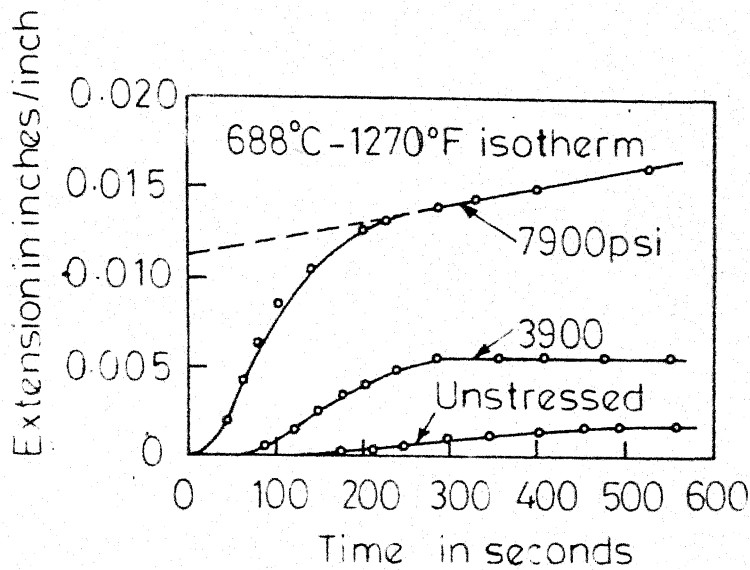


Fig. 3 - Extension measurements made during isothermal transformation to pearlite at 688°C. (2)

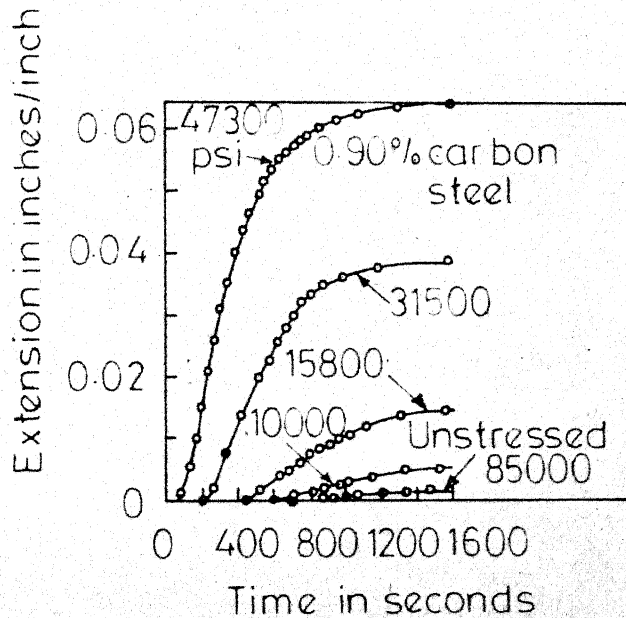


Fig. 4 - The progress of the deformation which occurs coincident with isothermal transformation to bainite at 260°C. (2)

Effect of stress on martensite transformation

No.136 nominal stress 984psi

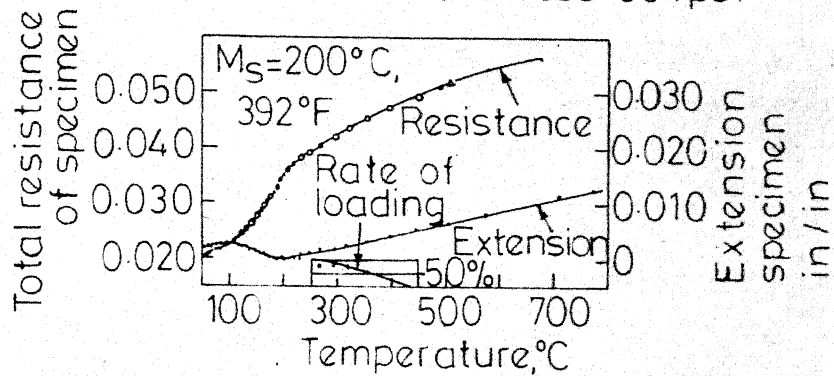


Fig. 5 -Results of resistance and extension measurements made during quenching to form martensite at low applied stresses. The lower plot gives the percentage of the total load applied in temperature range indicated.

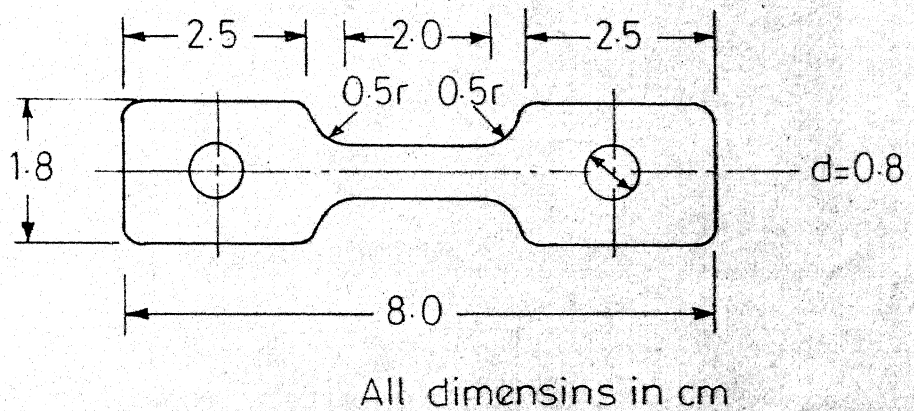
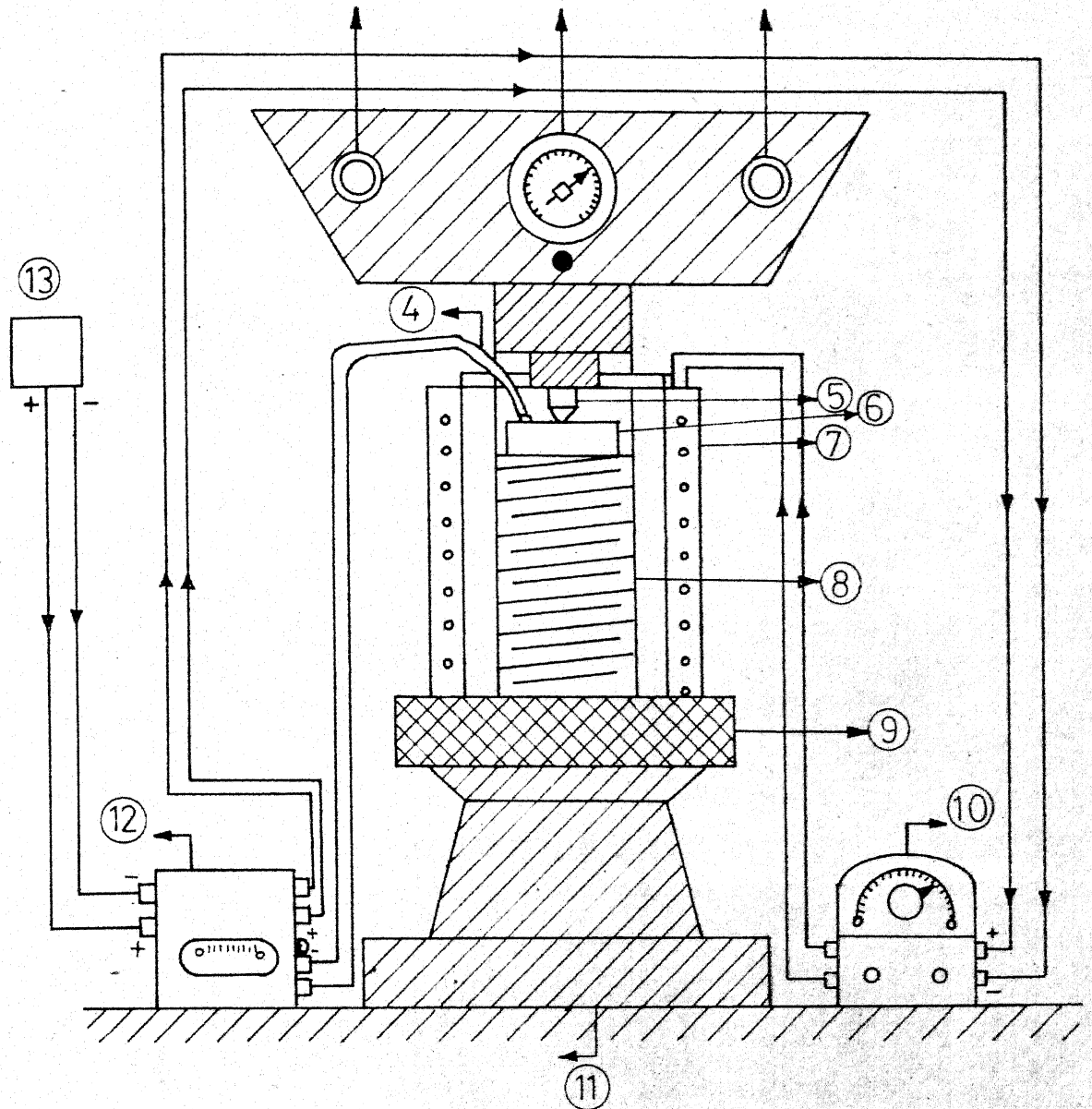


Fig. 6 -Tensile testing specimen.



1. Load releasing knob
2. Load dial indicator
3. Load applying knob
4. Alumel-chromel thermocouple
5. Indenter
6. Specimen
7. Furnace
8. Brinell stage moving screw
9. Brinell stage rotating plate
10. Dimmerstat (0-280V)
11. Brinell base
12. Temperature controller (0-1600°C) (Mahendra & Mahendra electro flow)
13. Mains -220V

Fig. 7 - Schematic diagram of hot hardness indentation test.

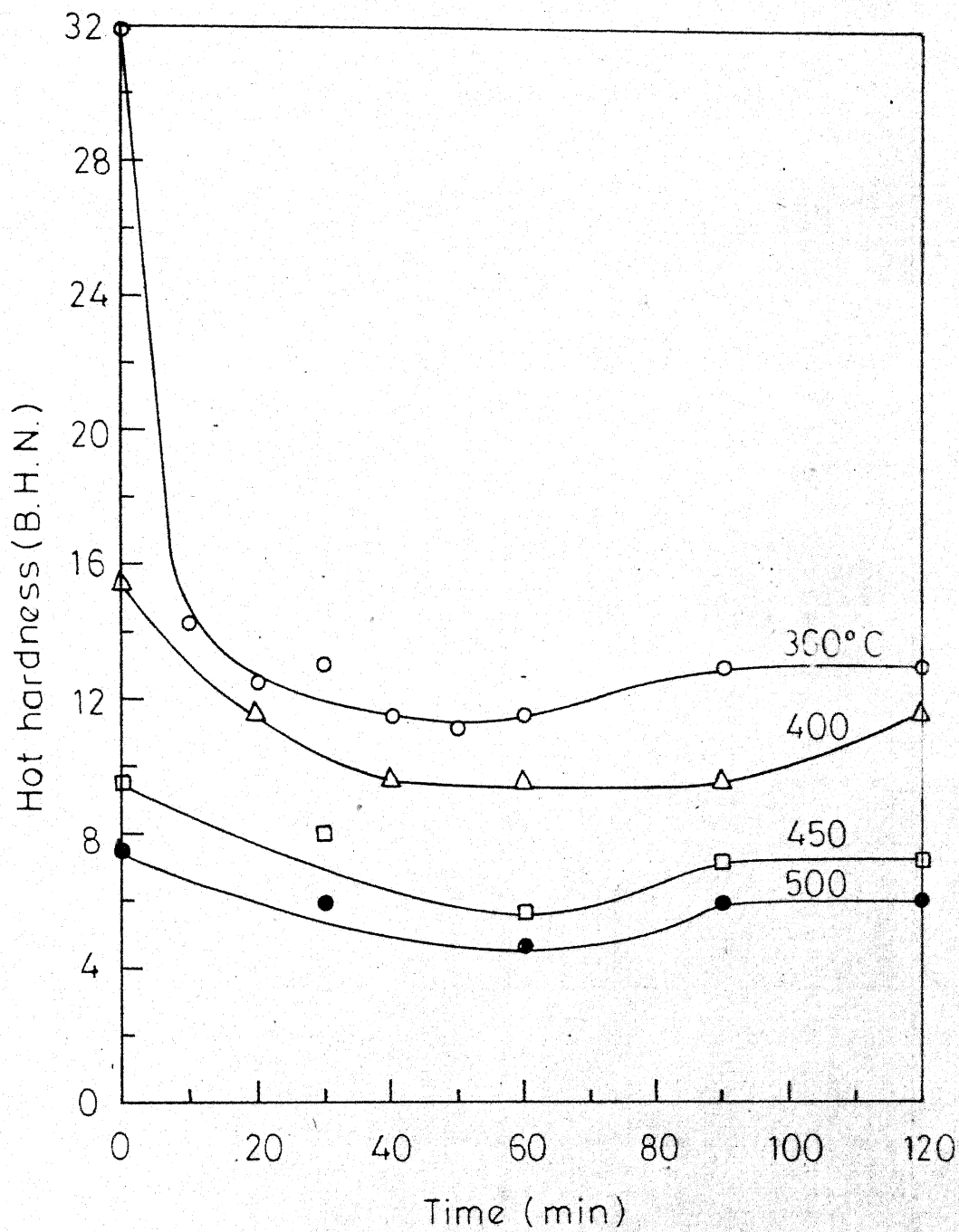


Fig. 8 -Effect of time on hot hardness of Al - 1% Cu alloy.

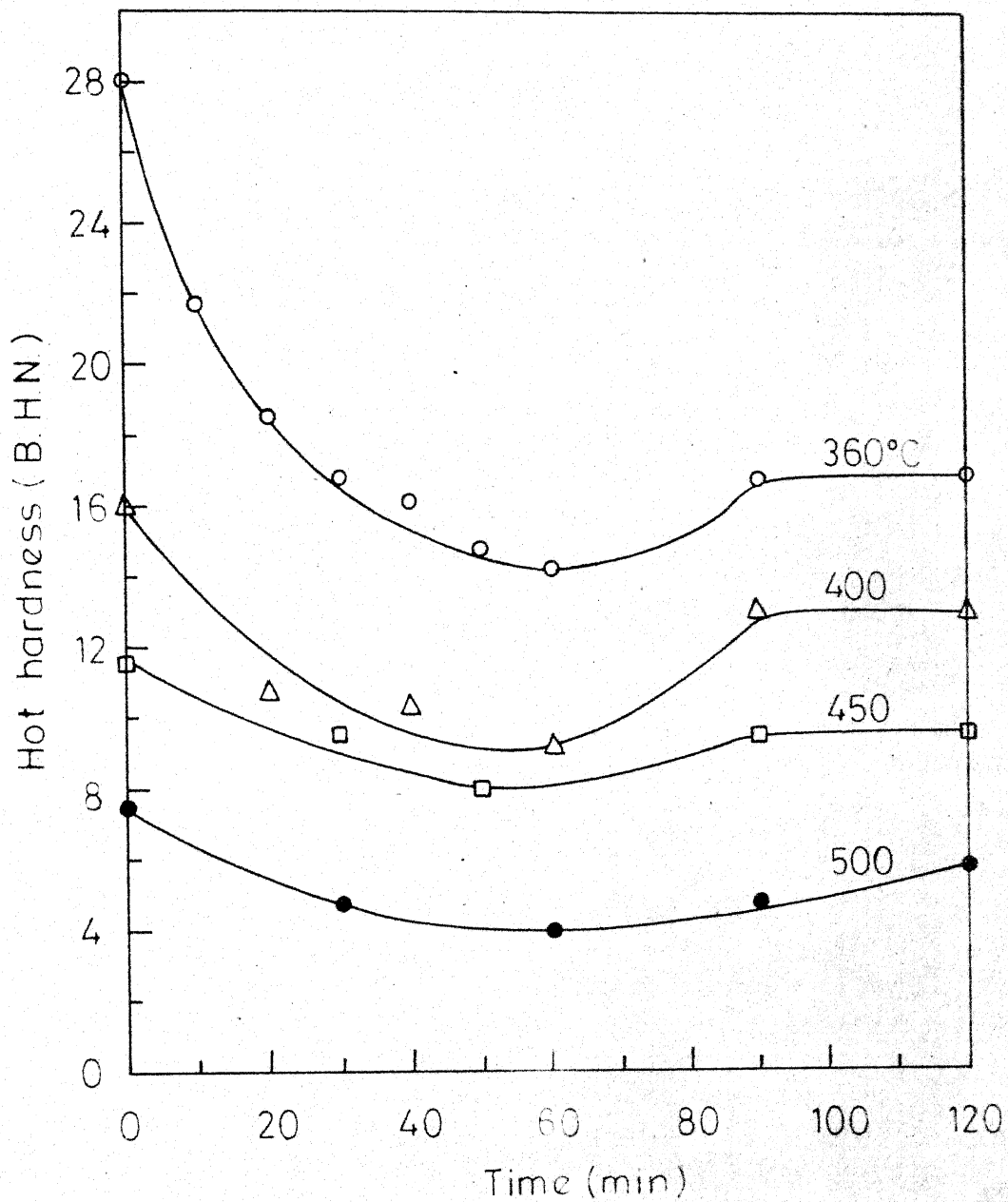


Fig. 9 -Effect of time on hot hardness of Al- 2% Cu alloy..

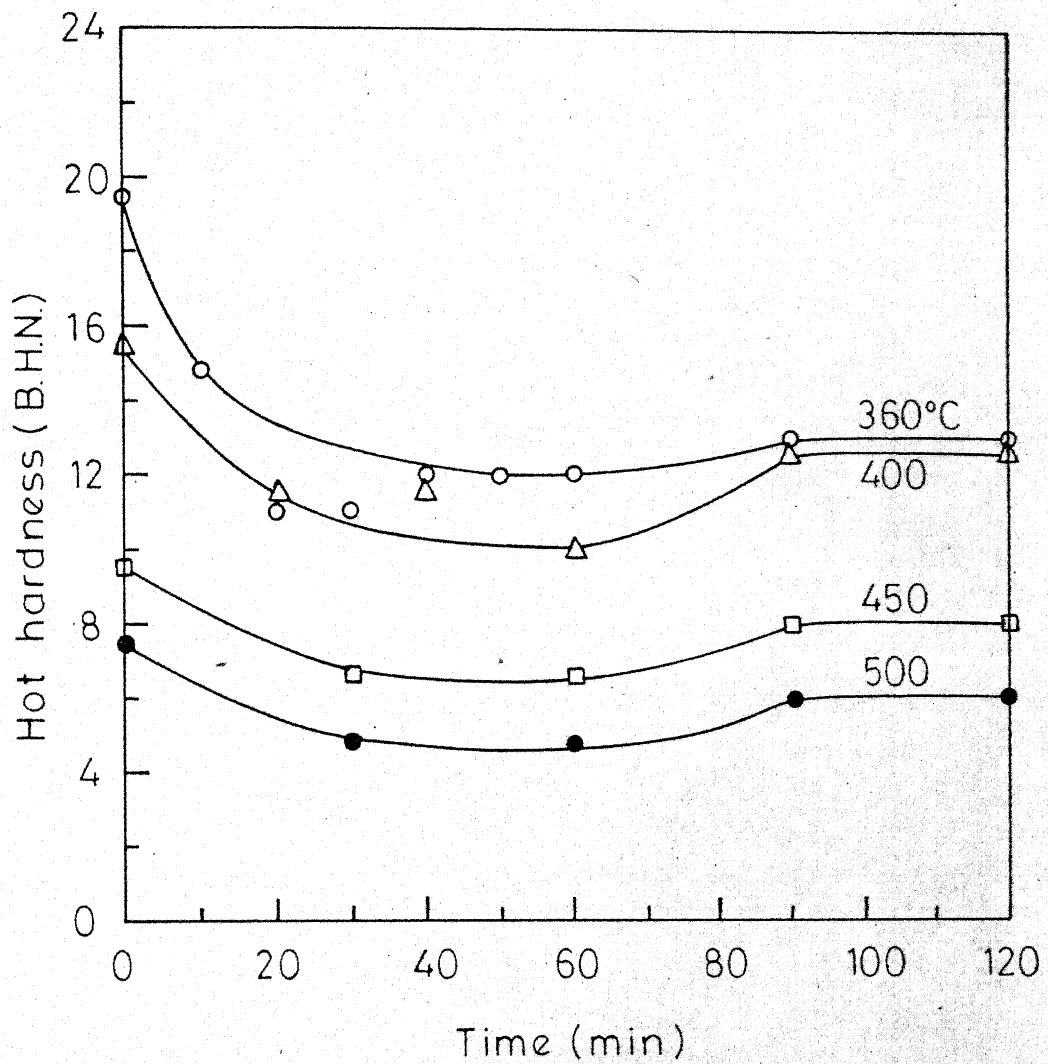


Fig. 10 - Effect of time on hot hardness of Al - 3% Cu alloy.

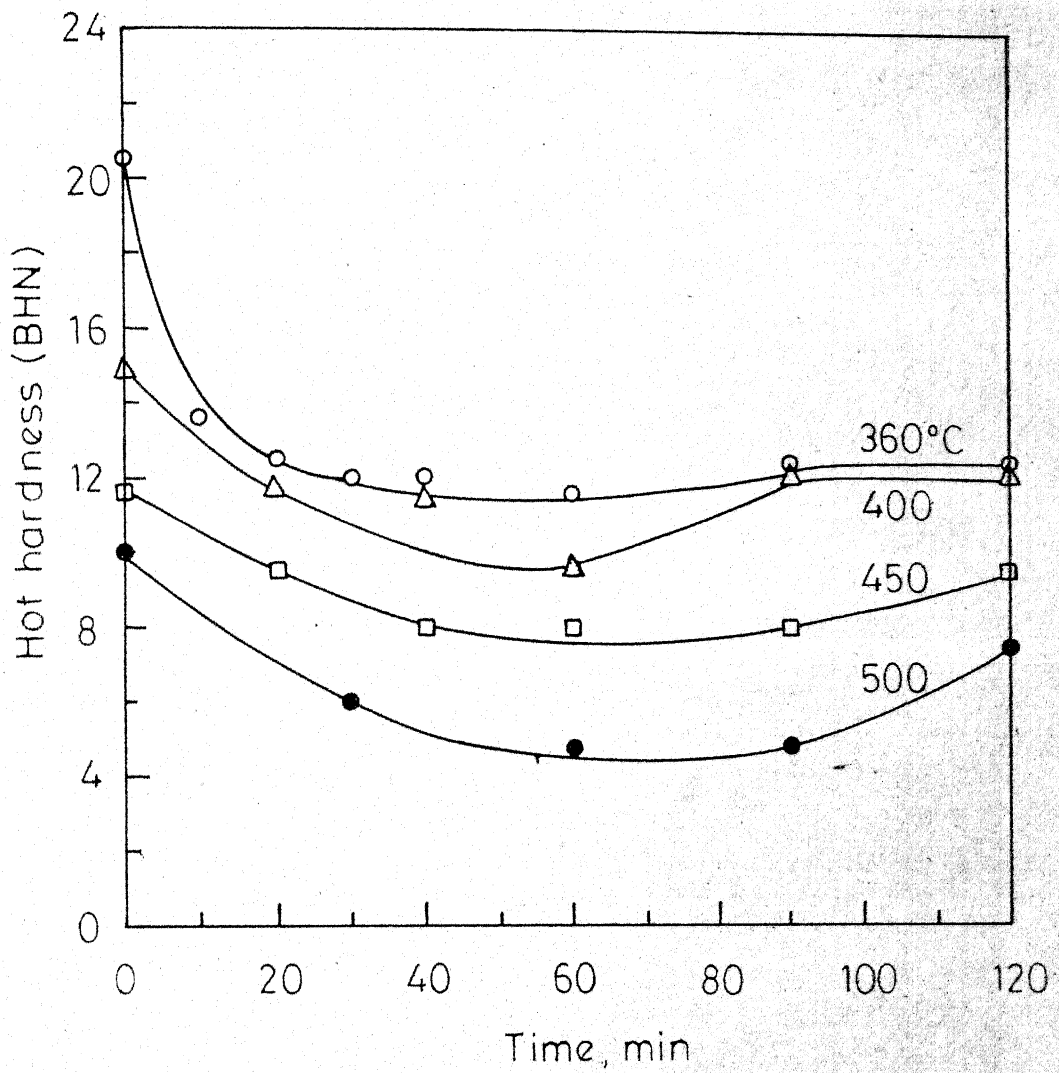


Fig. 11 - Effect of time on hot hardness of Al - 4% Cu alloy.

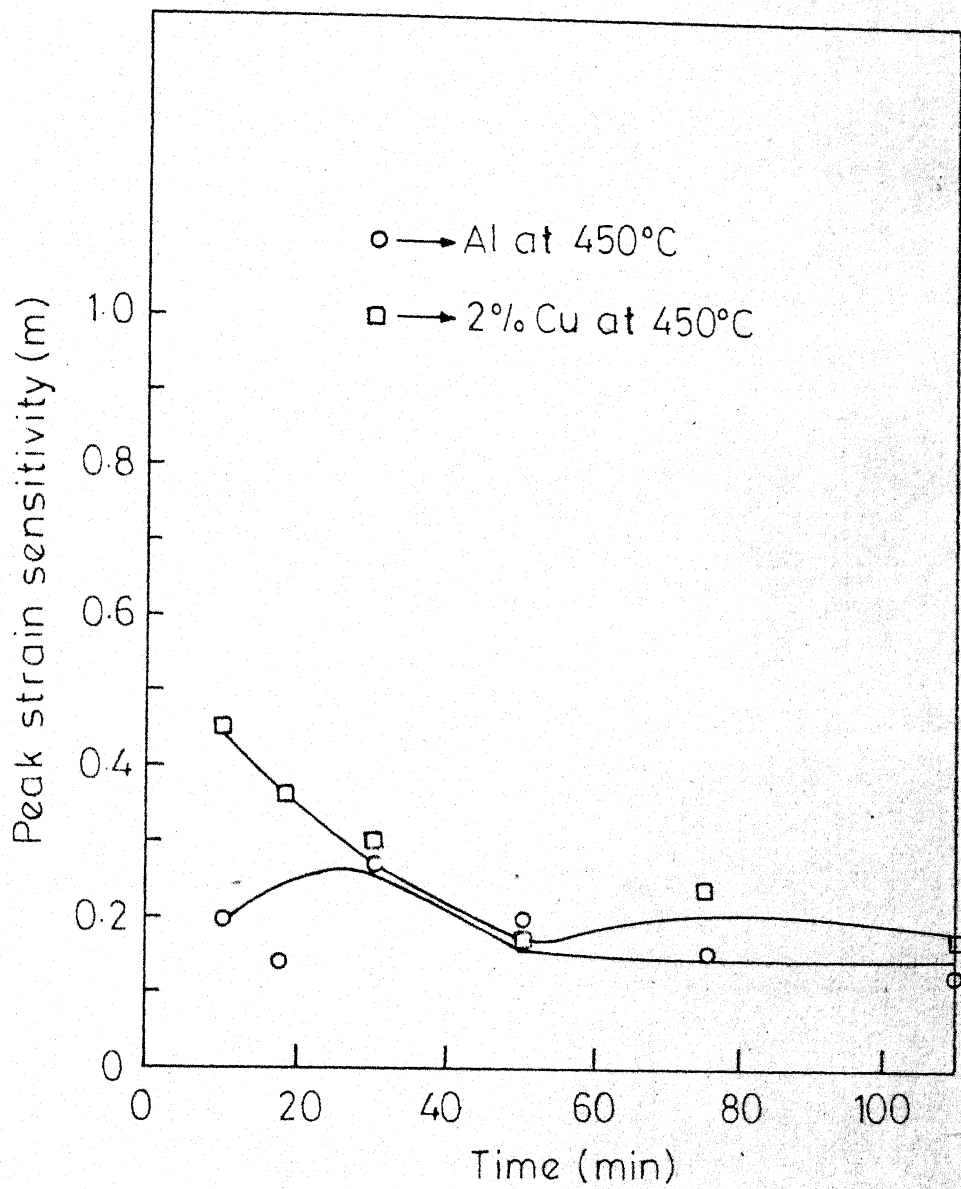


Fig. 12 - Comparison of peak strain sensitivity in Al-2% Cu alloy and commercial aluminium.

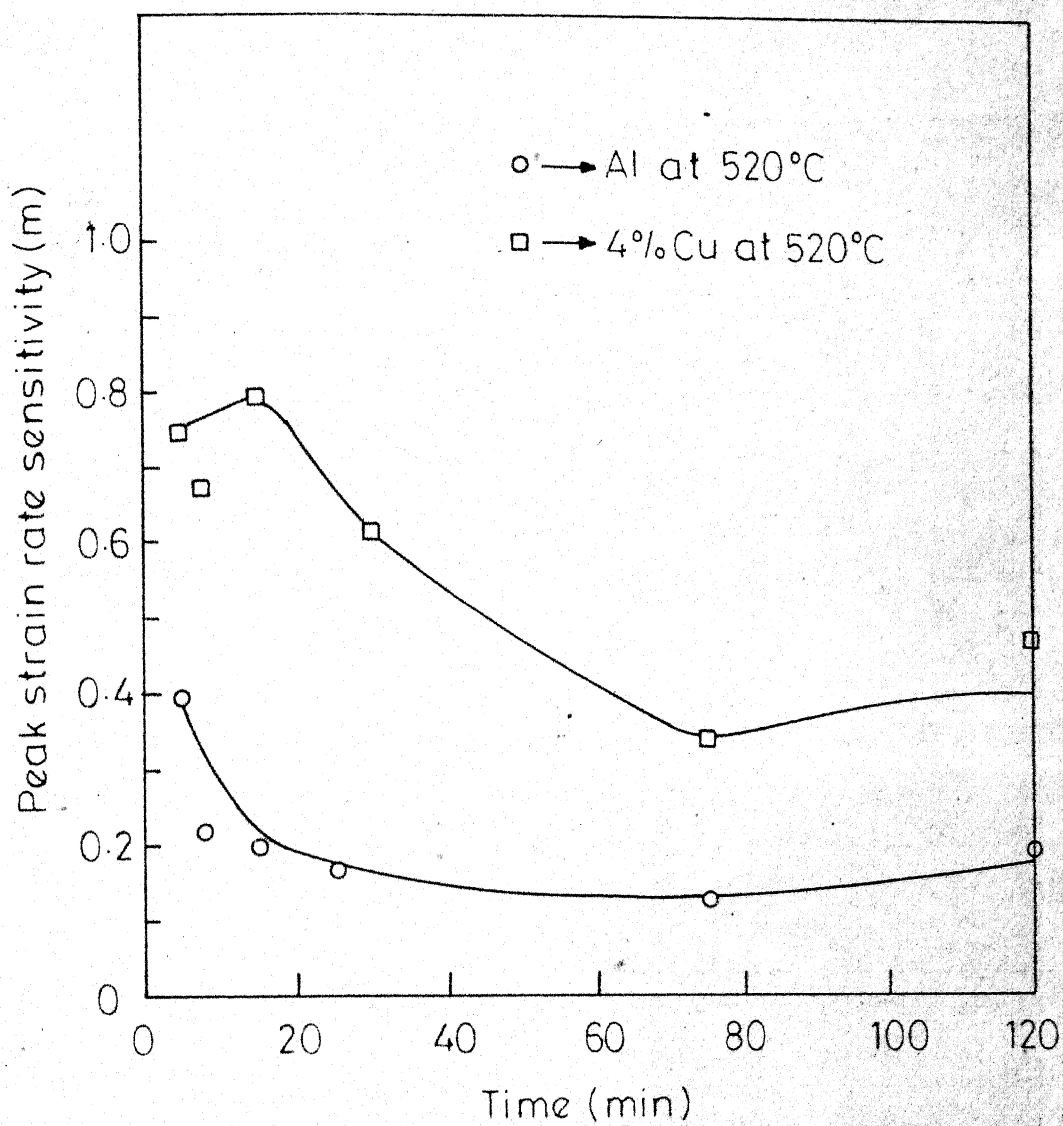


Fig 14 - Comparison of peak strain sensitivity in Al-4%Cu alloy and commercial aluminium.

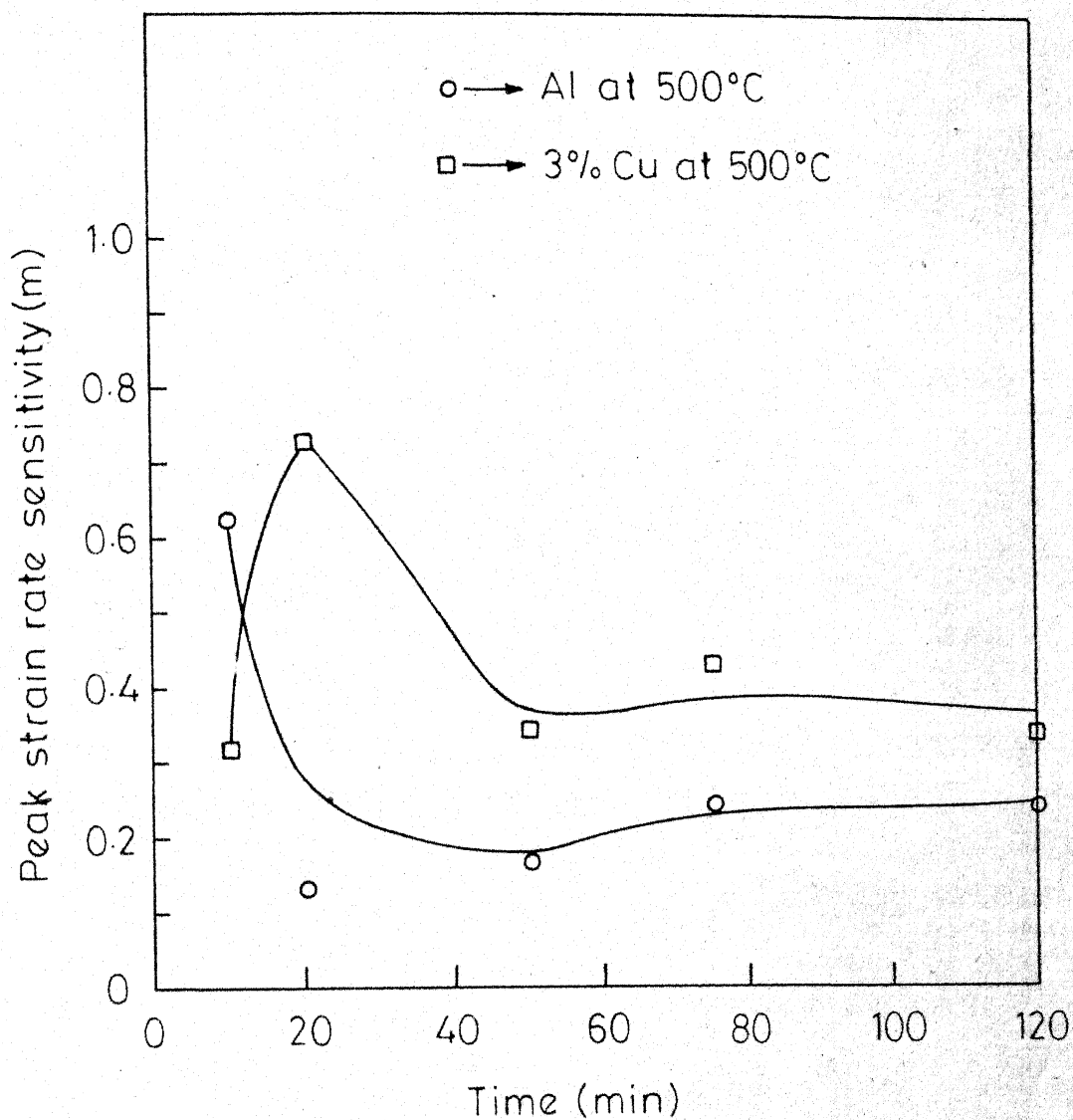


Fig. 13 - Comparison of peak strain rate sensitivity in Al-3%Cu alloy and commercial aluminium.

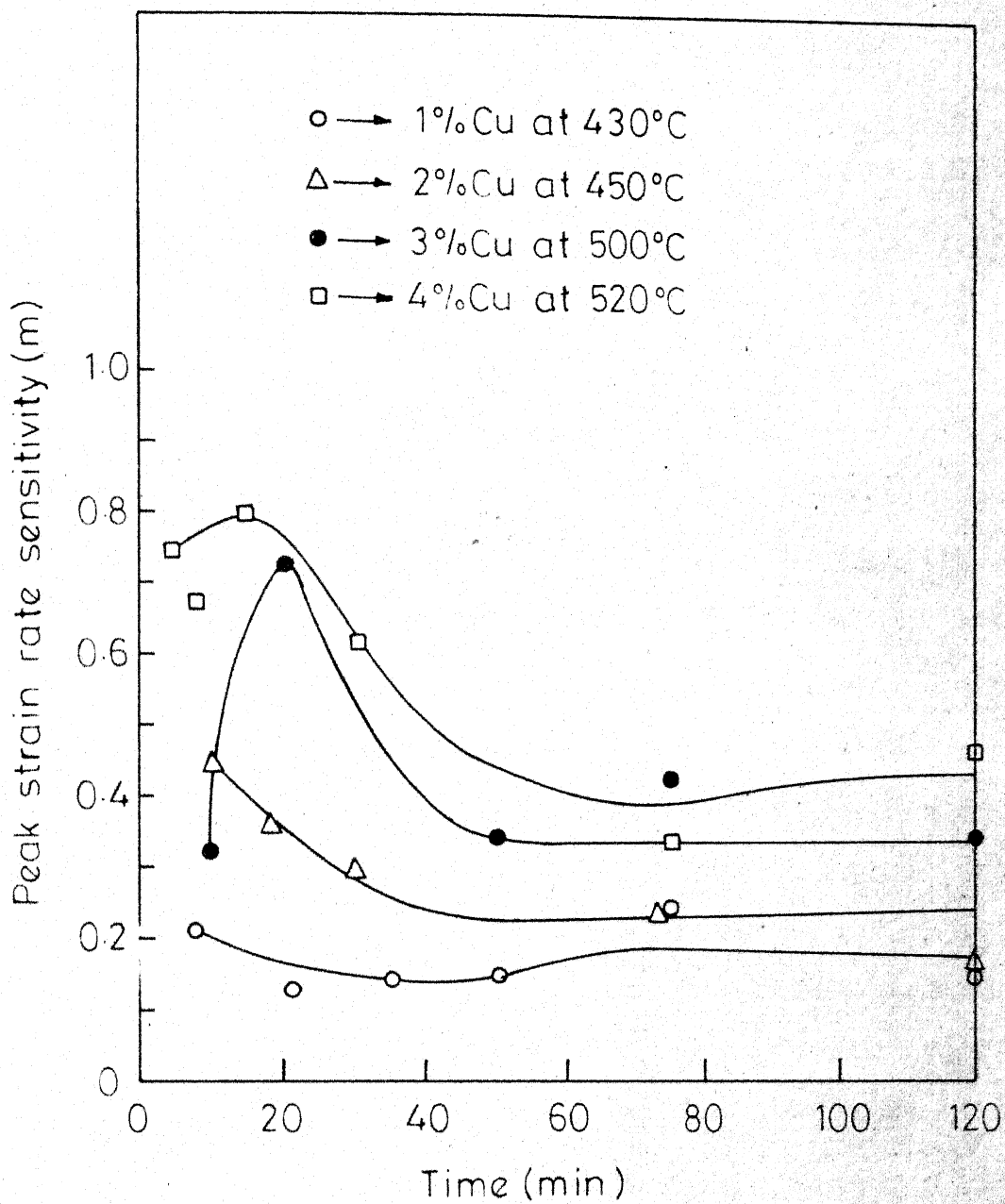


Fig. 15 - Effect of composition on peak strain sensitivity during $\alpha + \theta \rightarrow \alpha$ transformation.

CHAPTER - V

DISCUSSIONS

5.1 Indentation Tests

All the Al-Cu alloys contain two phases α and θ , at room temperature. The solid solubility of Cu in Al at room temperature is less than 0.1 percent. The solubility of Cu increases with temperature. Therefore, in any of the alloys tested, a dissolution reaction is expected to occur when heated above the room temperature. The quantity of copper that dissolves would depend upon the test temperature. The dissolution reaction would also cause change in volume fraction of the two phases.

The average hardness of an alloy depends upon (1) the strength of the individual phases (2) their volume fraction (3) test temperature. θ , being an intermetallic compound, is a strong phase as compared to α . The hardness is therefore expected to drop with drop in volume fraction of ' θ '. At any test temperature the volume fraction of ' θ ' changes with time due to the dissolution reaction hence, the hardness also decreases with time. Normally, the hardness would drop to a certain value over a period of time and then remain constant. But in all the alloys that were tested, hardness reaches a minima and again increases with time. This would be

possible only if additional softening processes such as Nabarro or coble creep flow is taking place.

Further, the reduction of hardness is spread over a period of nearly 1 hour which indicates that the concurrent reaction, $\alpha + \theta \rightarrow \alpha$ must be spreading over that period. It would sound unusual at the first instance but when one considers the process of homogenization as a part of the reaction, then such long reaction times as 1 hour would be reasonable and quite possible.

5.2 Comparison of Peak Strain Rate Sensitivity in Commercial Aluminium and Al-Cu Alloys

The difference in SRS values of commercial aluminium and Al-Cu alloys is attributable to the existence of concurrent transformation in the alloys as against its absence in the commercial aluminium. The concurrent reaction in the alloys helps sustain high SRS for a period higher than one hour since the homogenization reaction would continue to operate for such periods.

5.3 Variation of Peak Strain Rate Sensitivity at a Constant Temperature in Single Phase Region

The variation of peak SRS values for a given composition with increasing holding time can be explained in terms of transformation periods, and the volume fractions

involved in the transformation. As the holding time increases, the transformation would have been partially completed and less volume fraction would be available for transformation during the tensile test.

The volume fraction of an alloy involved in transformation process would depend on the composition of that alloy. With increasing copper percentages, the volume fraction going into the dissolution reaction would increase. It is this fact which would explain the variations in peak SRS values with time in different composition alloys. The alloy with the 4 percent Cu shows greater peak 'm' values as compared to those of 3 percent, 2 percent or 1 percent Cu alloy.

Further, the larger volume fractions involved in the transformation would take longer period for completion of the reaction. Hence higher peak 'm' values would be sustained for longer times in alloys containing 3 and 4 percent Cu.

5.4 Comparison of Peak Strain-Rate Sensitivity in Single and Two Phase Region:

The dependence of peak 'm' values on total volume fraction involved in the transformation process would also explain the differences observed in the SRS values in single and two phase regions. In case of alloys containing

1 percent and 2 percent copper the volume fractions available for transformation do not differ widely in single and two phase regions, and therefore SRS values do not differ widely. In case of 3 and 4 percent copper alloys the volume fractions available for transformation during two phase region is less than that available in single phase. Therefore the SRS values are higher in single phase region.

The SRS values lie in the so called superplastic range ($m > 0.3$) when the deformation is carried concurrent to transformation. To explain such high SRS values one would have to invoke such mechanisms as diffusional flow and grain boundary sliding. The diffusional flow mechanisms (N-H creep, and coble creep) would predict SRS values as high as 1 where as the grain boundary sliding mechanism (Ball Hutchinson Model) could predict SRS of the order of 0.5. The other high temperature deformation mechanisms predict lower SRS values.

For N-H creep, coble creep (Appendix II) or grain boundary sliding to operate at strain ranges (10^{-3} min^{-1} and above) a fine grain size in the range of 1 to 10μ must exist. It is proposed that the dissolution reaction in the Al-Cu system would transitorily lead to extremely fine grain size and make the above mechanism operative. This is the same mechanism as proposed by Agarwal and Vaidya [17]. The lowering of strength during the transformation would also indicate rapid diffusional flow there by supporting the above model.

CHAPTER - VI

CONCLUSIONS

The following conclusions are drawn from the present investigation:

- (1) In Al-Cu alloys softening occurs during the transformation reaction $\alpha + \theta \rightarrow \alpha$.
- (2) During the transformation the SRS values are far better in comparison with commercial aluminium.
- (3) During transformation the SRS values increase with percentage of Copper (1 - 4 percent Cu). As the holding time ranges upto two hours the SRS values remain above 0.3 in case of 3 and 4 percent Cu. In comparison with two phase region these values are better at higher percentages of copper.

PHENOMENOLOGICAL EXPLANATION OF SUPERPLASTIC BEHAVIOUR

It is necessary to examine briefly the mechanics of of stable flow under tensile forces in order to introduce the parameter of strain rate sensitivity (SRS). The explanation follows that given by Backofen et al [29]

A ductile material when stretched by a tensile force can deform uniformly only while stable flow occurs (homogeneous deformation). The limit of the stable flow is marked by the onset of geometrical instability-sudden concentration of deformation at point - called necking.

Necking generally occurs in a material elongated under tensile load. Because $V = A.l$ is fixed during plastic deformation, an incremental extension, δl , may be offset by incremental area reduction, δA , to insure

$$\begin{aligned} \delta V &= A\delta l + l\delta A = 0 \\ \text{or } \frac{\delta l}{l} &= \frac{\delta A}{A} = \delta \epsilon \end{aligned} \quad (1)$$

which defines an increment of true strain ϵ . The area reduction can have two effects.

1. It causes a loss in load carrying ability of an amount $\sigma\delta A$, where σ is the prevailing tensile flow stress when δA occurs. Rate of loss $= -\frac{\sigma\delta A}{\delta A} = -\sigma$

2. It may also cause increased load carrying capacity if strain hardening occurs, i.e. $\frac{d\sigma}{d\epsilon} > 0$

Now, to avoid microscopic necking, the incipient neck should not grow, so $-\frac{d\sigma}{d\epsilon} > \sigma$.

The usual 'exhaustion' of the strain hardening capacity in metals is responsible for a $\frac{d\sigma}{d\epsilon}$ to fall as ϵ increases and necking would begin when $\frac{d\sigma}{d\epsilon} = \sigma$.

In more general terms, σ is a function of many variables

$$\sigma = \sigma(\epsilon, \dot{\epsilon}, T, \gamma, \dots) \quad (2)$$

where ϵ is strain, $\dot{\epsilon}$ = strain rate, T = Absolute temperature of deformation and γ = surface energy.

Differentiating

$$\frac{d\sigma}{d\epsilon} = \frac{\partial \sigma}{\partial \epsilon} + \left(\frac{\partial \sigma}{\partial \dot{\epsilon}} \right) \left(\frac{d\dot{\epsilon}}{d\epsilon} \right) + \left(\frac{\partial \sigma}{\partial T} \right) \left(\frac{dT}{d\epsilon} \right) + \left(\frac{\partial \sigma}{\partial \gamma} \right) \left(\frac{d\gamma}{d\epsilon} \right) \quad (3)$$

This means that strain induced hardening to counteract neck formation may have multiple origins, metallurgical cold work (i.e., strain hardening), and increase in strain rate, a drop in temperature, or a rise in surface energy. The one to predominate determines many details of normal material behaviour.

At temperatures above $0.5 T_m$, as are the temperatures in superplastic phenomenon, contribution from $\frac{\partial \sigma}{\partial \dot{\epsilon}} \rightarrow 0$ and the second term in the equation (3) becomes more important; in other words the strain rate sensitivity ($\frac{\partial \sigma}{\partial \dot{\epsilon}}$).

A convenient $\sigma (\dot{\epsilon})$ relationship is $\sigma = K \dot{\epsilon}^m$, in which m is a very useful index of strain rate sensitivity given by $m = d \log \sigma / d \log \dot{\epsilon}$.

$$\text{Now, } \frac{P}{A} = \sigma = K \dot{\epsilon}^m \quad (4)$$

where P = load, A = Area of cross-section.

At any point along a rod in tension

$$\dot{\epsilon} = \frac{\delta l}{l} \frac{1}{t} = -\frac{1}{A} \frac{dA}{dt} \quad (5)$$

$$\text{combining (4) and (5) } -\frac{dA}{dt} = \left(\frac{P}{A} \right)^{1/m} \left[\frac{1}{A \frac{1-m}{m}} \right]$$

As long as m is < 1 , the smaller the cross-section the more rapidly it is reduced. As m approaches 1, the reduction rate at all cross-sections approaches a common level. When $m=1$, the flow is Newtonian viscous and $\frac{dA}{dt}$ is independent of A , so that any cross-sectional irregularities are simply preserved during pulling.

In a superplastic metal, where m is high but < 1 , this effect still exists. In a nonsuperplastic material in which m is always < 0.2 , a pronounced change

in cross section (neck) occurs where the area is a minimum and stress higher than elsewhere. Any local stress increase in the superplastic metal produces only small change in strain rate and consequently such regions tend to deform at a rate not significantly different from the rest of the specimen and a very much extended version of a neck results. This effect makes it appear almost as if the specimen was not necking at all.

APPENDIX - II

There are many well established high temperature structural mechanisms proposed in the context of high temperature steady state creep phenomenon and hot working processes. Some of them have been invoked to explain the superplastic behaviour of various workers. It is therefore, pertinent to know details about the relevant mechanisms.

These mechanisms could be classified into two groups (i) those operative at low stresses (ii) those operative at high stresses.

The various low stress high temperature diffusion controlled mechanisms that have been identified as

- Atomic transport processes (1) Nabarro-herring mechanisms
- (2) Coble mechanisms

Grain boundary sliding mechanisms

The various high stress mechanisms that have been identified are

Dynamic Recovery Process

Dynamic Recrystallization Process

Atomic Transport Processes

At sufficiently high temperatures ($\geq 0.5 T_m$) and low stresses crystalline solids can deform at a significant rate by atomic transport processes. This phenomenon is

possible because there exists, in real crystalline materials, locations where atoms can be added to, or removed from, the crystal lattice. When the material is stress free, and in equilibrium, the chemical potentials at these sinks and sources are everywhere the same, but on application of stress, providing that is not purely hydrostatic, chemical differences arise. These provide a driving force for diffusion between sources and sinks, and the resultant transport of matter changes the shape of the material in response to the deforming forces. The sources and sinks for atoms may be free surfaces, grain boundaries or dislocations, while diffusion paths may be through the lattice, through a vapor phase, along a free surface, along grain boundaries or along dislocation cores.

Nabarro-Herring Creep Mechanism [30,31]

Nabarro-herring postulated that the creep results from the diffusion of vacancies from the regions of high chemical potential at grain boundaries subjected to normal tensile stresses to regions of lower chemical potential where the average tensile stresses across the grain boundaries are zero. Atoms migrating in the opposite direction through the lattice account for the creep strain. When the volume diffusion controls the steady state tensile creep rate, is given by

$$\dot{\epsilon} = A_n \cdot \frac{D_v b^3 \sigma}{d^2 kT}$$

where b is the burger's vector, σ is the applied stress, d is the mean grain diameter and kT is the Boltzmann constant times the absolute temperature. A_n is a constant. The diffusivity D_v , is obtained from the tracer diffusivity D . For pure metals

$$D_v = \frac{D^*}{f}$$

and for binary solution

$$D_v = \frac{D_A^* D_B^*}{(N_A D_A^* + N_B D_B^*) f} \quad (3.3)$$

where N_A and N_B are the atomic fractions of A and B atoms, and f is the correlation factor. The dimensionless constant A_n depends insensitivity on the geometry of grains, but is generally estimated to have a value from 8 to 5.

Nabarro creep does not invoke the motion of dislocation. It predominates over high temperature dislocation dependent mechanisms only at low stress levels and then, only for fine grained materials.

In order to provide a ready comparison of Nabarro creep with other mechanisms to be discussed later, the equation (3.1) is reformulated as

$$\frac{\dot{\epsilon} kT}{D_v Gb} = A_n \left(\frac{b}{d} \right)^2 \left(\frac{\sigma}{G} \right)^1 \quad (3.4)$$

where G is the shear modulus of elasticity.

Coble Creep Mechanism [32]

Coble creep results from the diffusion of vacancies from regions of high chemical potential at grain boundaries subjected to normal tensile stresses to regions of lower chemical potential where average tensile stress across the grain boundaries is zero. Atoms migrate via grain boundaries instead of migrating through the lattice as in the case of Nabarro creep. The creep is given by an expression similar to that for Nabarro creep except for the grain size dependence. The equation as reformulated is

$$\frac{\dot{\epsilon} kT}{D_{gb} G b} = A_n' \left(\frac{b}{d} \right)^3 \left(\frac{\sigma}{G} \right)^1 \quad (3.5)$$

where D_{gb} is the grain boundary diffusivity and A_n' is a constant which depends insensitively on the geometry of grains and estimated to have value of 48.

The grain boundary diffusion mechanism should predominate over volume diffusion when

$$48 \left(\frac{b}{d} \right) \left(\frac{D_{gb}}{D_v} \right) > 7 \quad (3.6)$$

Theories based on Grain Boundary Sliding (GBS)

Gifkin^[33] was the first to propose a model based on GBS. According to him the strain rate over the

superplastic region was given by

$$\dot{\epsilon} = B_3 \frac{\sigma \lambda}{kT} D_{gb}$$

where $B_3 = \frac{\text{constant}}{d^3}$. The expression gives a d^{-3} dependence and a stress dependence of unity of the strain rate which conflicts with some of the observations.

Ball and Hutchison [34] proposed a model where the strain rate was mainly given by GBS where dislocation-climb controlled mechanism acted as an accommodation process. They argued that groups suitably aligned grain slide over each other. Certain grains that obstruct the easy relative motion yield under resulting stress concentration. Under superplastic conditions, dislocations traverse such yielded grains and pile up at grain boundaries until their back stress prevents the boundary sliding. The high stress at the head of the pile up causes accelerated diffusion and dislocations rapidly escape by climb into and along grain boundaries. The strain rate is given by an expression

$$\dot{\epsilon} = \frac{K\sigma^2}{d^2} \exp(-U/kT)$$

where U = activation energy for grain boundary diffusion and k is a constant.

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